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APPLICATION OF FRACTURE MECHANICS TECHNIQUES TO HIGH TEMPERATUR--ETC(U)

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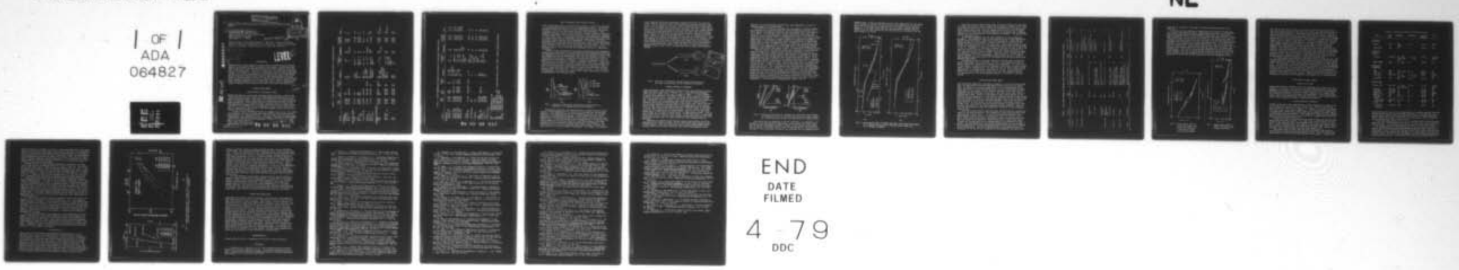

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APPLICATION OF FRACTURE MECHANICS TECHNIQUES TO HIGH TEMPERATURE CRACK GROWTH.

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INTRODUCTION

There has been a large impetus in recent years to extend fracture mechanics concepts to characterize high temperature crack growth. Several years ago McEvily and Wells[1] reviewed the application of these concepts to safe design in the creep range. Since then a large body of information on high temperature crack growth in a variety of structural alloys has been accumulated and there is a better understanding of the factors that affect crack growth and the limitations these factors impose on the characterization of crack growth by fracture mechanics concepts. Based on the available information we shall attempt to summarize here the applicability of fracture mechanics techniques to high temperature crack growth under cyclic, static, and combined loads and emphasize the conditions that could limit their applicability.

FATIGUE CRACK GROWTH

Crack Growth in Structural Alloys

Recognition that the life of many thermostructural components could be limited by subcritical crack growth under high temperature fatigue has led to many crack growth studies. The success of linear elastic fracture mechanics (LEFM) to characterize low temperature crack growth has motivated its use at high temperatures. Crack growth behavior in many structural alloys under high temperature fatigue [2-35] has been studied and these have been listed in Table 1. These extend from low temperature alloys, such as aluminum alloys, to superalloys, and to some of the more advanced alloys such as directionally solidified eutectics.

In most of these investigations LEFM is assumed to be valid. Only a few investigations were done to check their applicability using different specimen geometries and loading conditions. For cases where LEFM was found to be inapplicable, other parameters such as J-integral, crack opening displacement and net section stress have been suggested. Before discussing these parameters it is useful first to r[illegible] that affect high temperature crack [illegible] 251 950 [illegible] n the applicability of fracture mec[illegible]

Table 1 - High Temperature Fatigue Crack Growth in Structural Alloys

Alloy	Temp (°C)	Y.S. (MPa)	Tensile (ε%)	Geometry	Frequency (Hz)	Parameter	ΔK Range (MPa$\sqrt{m}$)	Ref
Al-Alloys								
7075-7651	24-100	503		CC	143	ΔK	6-20	2
Al-2.6 Mg	24-300	-	-	WOL	1	ΔK	2-20	3
RR-58	20-190	392-310	-	CC	5	ΔK, $\Delta K/E$	7-40	4
Low Alloy Steels								
A212B	24-510	275		CT, CC Surface	.002 to .1	ΔK	20-100	5,6
A517F	24-510	689		CT	.002 to .1	ΔK	20-100	5
9Ni, 4Co, .3C	24-510	1309		CT	.002 to .1	ΔK	20-100	5
2 1/2 Cr-1Mo	24-593	220-130		WOL, CT CC, Surface	.06 - 6.6	ΔK	5-40	6,7
Cr-Mo-V (Air, Vac.)	24-550	600-300	20-30	WOL, SENT CC	$10 - 10^{-3}$	ΔK, COD σ_{net}	5-40	8,9
Cr-Mo-V (Air, Vac.)	24-565	670-300	36	SEN-Tension Bending	Hold-Time Static	ΔK	30-200	10
Austenitic Stainless								
A286	24-538	689-769	23-20	CT	3	ΔK	12-60	11
304	24-650	273	65	WOL, SEN, CT, Round bar Double edge	.17 - 6.6	ΔK, σ_{net}	10-50	6,12-14
308 weld	24-704	427-283	66	SEN, CT	.17 - 11.2	ΔK	10-80	15
316 sol, c.w.	24-593	304-165	57	SEN, CT	.17 - 6.6 Hold-Time Static	ΔK	10-80	12,13,16
321	24-593	213-193	57	SEN	.17	ΔK	10-50	13
348	24-593	241-220	51	SEN	.17	ΔK	10-50	13

Table 1 - High Temperature Fatigue Crack Growth in Structural Alloys (Cont)

Alloy	Temp (°C)	Y.S. (MPa)	Tensile (ε%)	Geometry	Frequency (Hz)	Parameter	ΔK Range (MPa√m)	Ref
Superalloy								
Incoloy 800	24-704	1038	53	CT, SEN	.17	ΔK	15-90	17,18
Alloy 718	24-760	1100	15	CT, SEN CC, Surface	.17 - 66	ΔK	15-90	19-23
Inconel 600	24-649	243-165	31-17	CT, SEN	.0014-6.6	ΔK	15-90	18,24
Inconel 625	24-704	-	-	SEN	.17	ΔK	15-90	18
Inconel X750	24-704	648-541	14-5	CT, SEN	.17, 0.66	ΔK	15-90	18,25
Nimonic PE16	24-649	620-570	20-28	CC, SEN	.17, 0.66	ΔK	10-70	18,26
Udimet 700	24-850	-	-	CT	.17 Hold-Time Static	ΔK, ΔJ COD	10-80	27-29
IN 100 Gatorized	24-734	1130-1075	14-22	CT, CC	0.66 Hold Time	ΔK	10-80	30
P/M Astroloy	650	-	-	CT	1 - 0.01 Static	ΔK	40-70	31
Waspaloy	650	-	-	CT	1 - 0.01 Static	ΔK	40-70	31
HS25	24-704	-	-	SEN	.17	ΔK	15-90	18
Haynes 188	24-870	430-213	-	DCB	.01 - 10	ΔK	30-90	32
IN 738	24-850	-	-	DCB	100, Static	ΔK	7-60	33
γ-γ'/δ DS	24-1038	-	-	CC	.17	ΔK	10-80	34
C-73 (Co,Cr)-Cr_7C_3	25-950	-	-	CC	-	ΔK	4-50	35

WOL - Wedge opening load
CC - Center Cracked
CT - Compact Tension
SEN - Single edge notch, SENT - Single Edge Notch Tensile, SENB - Single Edge Notch Bending
DCB - Double Cantilever Beam
DEN - Double Edge Notch
Surface - Surface Cracked

Crack growth under cyclic load can occur by either or both of two processes: (1) cycle-dependent process, and (2) time-dependent process. The relative contribution from each process to crack growth depends on the temperature, frequency, hold-time or wave shape, material, and environment. Time-dependent processes include both creep and environmental effects, which are thermally activated, and thus could introduce a large temperature dependence in crack growth rates. On the other hand, purely cycle-dependent processes are rather insensitive to temperature, and any temperature dependence is due to elastic modulus variation with temperature. For example, the cycle-dependent process could be plastic blunting [36], where irreversible plastic flow under cycling provides the driving force for crack growth. Since initiation of plastic flow (dislocation nucleation) from the crack tip is an athermal process, we can treat the cycle-dependent process as temperature-independent.

Figure 1 shows schematically the effect of temperature and frequency on fatigue crack growth rates for time-dependent and cycle-dependent processes. Low frequencies and high temperatures favor the time-dependent process, while high frequencies and low temperatures favor the cycle-dependent process. At intermediate temperatures, a combination of both processes could occur depending on the amplitude and frequency. Temperature dependence similar to that in Fig. 1a was observed in Type 304 stainless steel [12], a cobalt-base alloy [32] and in several nickel-base alloys [18]. Crack growth rates as a function of frequency was discussed earlier [37] with reference to A286 alloy. In some cases ultra high frequencies may be required before fatigue becomes purely cycle dependent. It is apparent from Fig. 1 that the effect of temperature on fatigue crack growth in the creep range cannot be studied independently of the frequency effect.

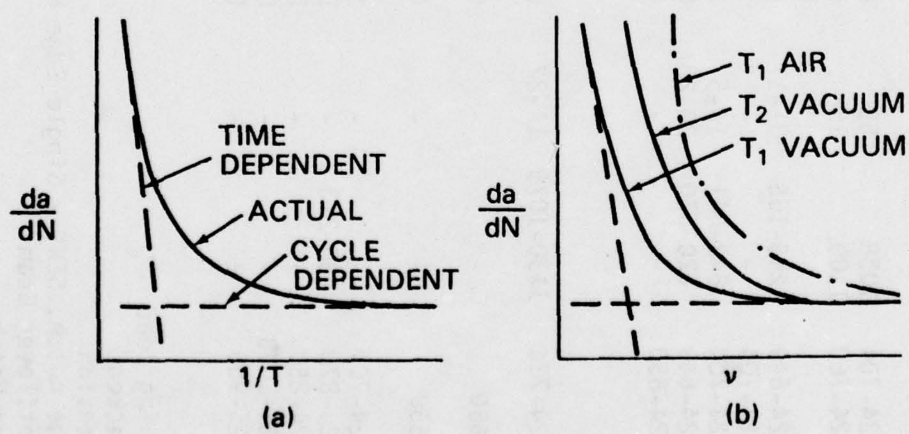


Fig. 1 Schematic illustration showing crack growth rate as a function of (a) temperature and (b) frequency

The question concerning the applicability of fracture mechanics techniques to high temperature crack growth can be divided into two separate questions: the applicability to a cycle-dependent process; and the applicability to time-dependent processes. If crack growth is purely time-dependent, cycling may not be necessary since cracks could grow even under static load. In the intermediate range of temperatures or frequencies, generally between 0.4 to 0.7 T_M, where T_M is the melting point, both cycle-dependent and time-dependent processes could occur simultaneously. Any interaction between the two processes could significantly influence the applicability of fracture mechanics techniques. For example, the enhanced

creep component during high temperature fatigue could alter the crack tip stress field to the extent that fracture mechanics techniques cannot be applied. This can happen in two ways. Stress relaxation due to creep could occur at a rate faster than the increase in stress field due to crack growth. This results in crack tip blunting which influences the crack growth process. Secondly, the creep deformation at the crack tip could cause the formation of fissures around the crack tip and reduce the stress concentration effect there. These are represented schematically in Fig. 2. Aggressive environments could also cause a similar type of crack tip stress relaxations. Crack tip blunting by corrosion or formation of fissures due to the formation of grain boundary brittle phases could both occur under severe environments. In the following we shall first discuss the applicability of fracture mechanics to cycle-dependent process, then the effects of superimposed time-dependent process, and finally, the applicability to purely time-dependent process.

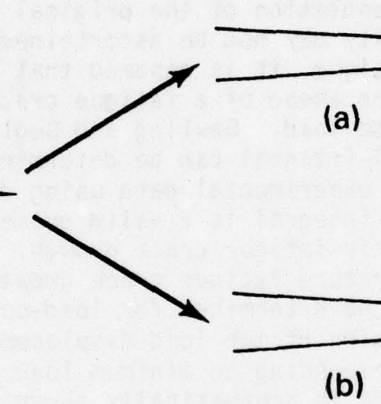


Fig. 2 Schematic illustration showing stress relaxation due to (a) crack tip blunting, and (b) formation of fissures

Fracture Mechanics Parameters

Crack growth under high temperature fatigue has been characterized by stress intensity factor range (ΔK), J-integral (ΔJ), crack opening displacement (COD), and net section stress (σ_N), as shown in Table 1. If crack growth is purely cycle-dependent, the stress intensity factor could be a reasonable parameter if plane strain conditions can be maintained ahead of the crack tip. Correlation with ΔK in most of the alloys was found at high frequencies and at relatively low temperatures where crack growth is mostly cycle-dependent. Note that the ASTM thickness criterion for plane strain [38] cannot be simply extended to fatigue even at room temperature, since cycling localizes the crack tip deformation [39] and extends the range of thickness or ΔK for which plane strain conditions exist. In some alloys at high temperatures, the superposition of environmental effects could increase crack growth rates so that stress relaxation due to creep becomes minimal. This in turn extends the ΔK range for plane strain. On the other hand, in some other alloys where plastic flow can occur rapidly, LEFM may not provide an accurate description of the crack tip stress fields and other parameters may be necessary.

Haigh [9] has recently evaluated the range of applicability of ΔK for tempered 1Cr-Mo-V steel at 550°C. LEFM is valid until gross plasticity due to creep occurs. The stress intensity limit at which creep becomes sufficiently significant to affect the crack tip stress fields depends on the

material, its creep strength and ductility, test temperature, as well as specimen geometry and loading mode.

Correlation of crack growth rates with $\Delta K/E$ where E is Young's modulus was obtained for an aluminum alloy [4] in the temperature range of 20 to 190°C and for an austenitic stainless steel weld metal [14] in the range of 24 to 704°C. This indicates that, for the temperatures used, the crack growth is mostly cycle-dependent for these alloys, and it is relatively insensitive to environment. Also, no differences in crack growth rates at room temperatures were observed in several nickel-base alloys that encompass a broad range in yield stresses while significant differences in their crack growth rates at 593°C were observed [18]. In addition, no correlation was obtained on $\Delta K/E$ basis at 593°C. Both imply that crack growth at room temperature is cycle-dependent, while at 593°C there is a superimposed time-dependent crack growth, the extent of which is different for different alloys.

When the plastic flow becomes significant, especially at high temperatures, LEFM may be of limited use and non-linear parameters such as J-integral or COD may be better suited. Application of J-integral to fatigue crack growth involves extrapolation of the original concept [40] to cyclic conditions where its validity may not be ascertained by the theory. In extending this concept to fatigue, it is assumed that there exists a characteristic elastic-plastic region ahead of a fatigue crack which is a unique function of the crack length and load. Dowling and Begley [41] have shown that, in A533-B steel, a proper J-integral can be determined under displacement-controlled fatigue. Their experimental data using different specimen geometries [52] suggests that J-integral is a valid geometry and size independent parameter for elastic-plastic fatigue crack growth. It has been shown recently [28] for high temperature fatigue crack growth in Udimet 700 that a proper J-integral can also be determined for load-controlled fatigue. The procedure involves integration of the load-displacement curves (rising load), taking the extensions corresponding to minimum load as the lower limit of integration. This procedure is schematically shown in Fig. 3.

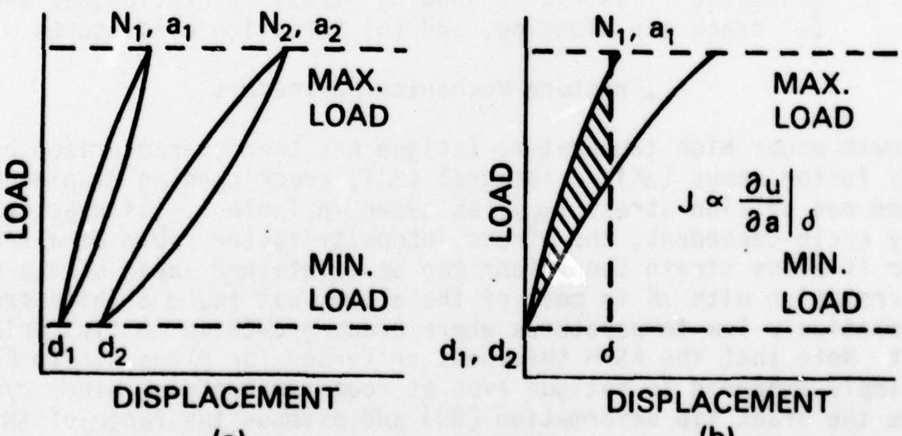


Fig. 3 Method of evaluation of J-integral for load-controlled fatigue; (a) hysteresis loops for two different cycles and crack lengths, (b) rising load parts of the loops displaced to a common origin.

This definition is compatible to the ΔK definition, and therefore it is possible to see the effects of mean stress, or the effects of stress ratio, similar to those observed on ΔK basis. Figure 4 shows fatigue crack growth data of Udimet 700 at 850°C [28] in terms of both ΔK and ΔJ. The spread in the data is less pronounced in terms of ΔJ, although, for the compact tension

geometry used. ΔK may be reasonably valid at this temperature for this alloy. It is possible to estimate ΔJ using a single load-extension curve and the Merkle-Corten estimation procedure [43] for the compact tension specimen geometry. It has been shown [28] that the estimation procedure gives ΔJ values close to those observed by the compliance-type of procedure for the whole range of ΔK values.

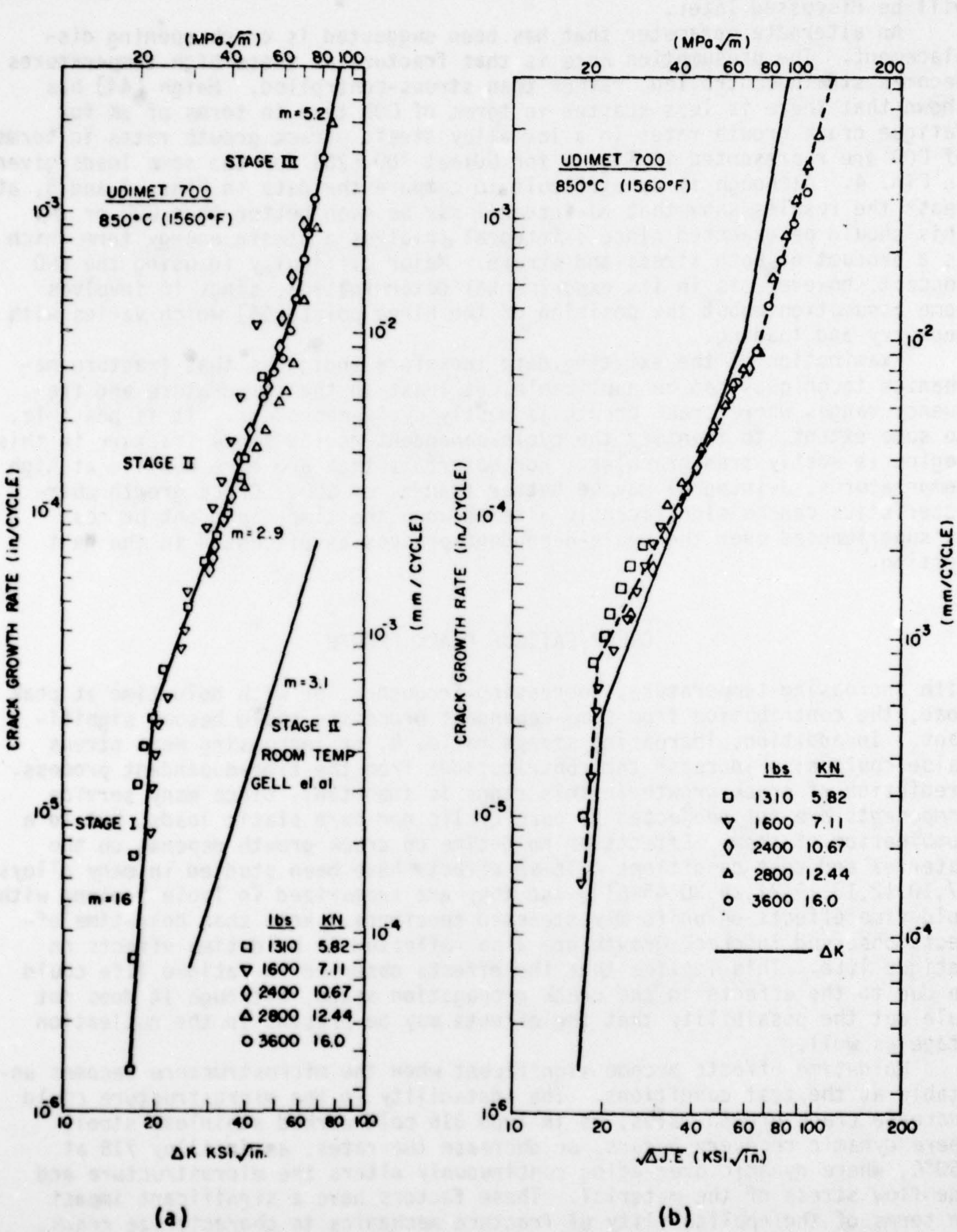


Fig. 4 Crack growth rates in Udimet 700 under load-controlled fatigue as a function of (a) stress intensity factor range, and (b) J-integral parameter

Note that unlike in A533-B alloy [41], the Stage II-Stage III transition in fatigue crack growth rates in Udimet 700 exists even in terms of ΔJ. This transition for this alloy occurs due to the superposition of a creep-related crack growth process [28]. Since ΔJ appears to be valid in this ΔK range, (see Fig. 4) it raises the interesting question as to whether the J-integral concept can also be extended to creep-induced crack growth. This possibility will be discussed later.

An alternate parameter that has been suggested is crack opening displacement. The presumption here is that fracture at these high temperatures becomes strain-controlled, rather than stress-controlled. Haigh [44] has shown that there is less scatter in terms of COD than in terms of ΔK for fatigue crack growth rates in a low alloy steel. Crack growth rates in terms of COD are represented in Fig. 5 for Udimet 700 [28] for the same loads given in Fig. 4. Although it is difficult to compare the data in Figs. 4 and 5, at least the results show that ΔJ-integral may be even better than COD or ΔK. This should be expected since J-integral involves a strain energy term which is a product of both stress and strain. Major difficulty in using the COD concept, however, is in its experimental determination, since it involves some assumption about the position of the hinge point [44] which varies with geometry and loading.

Examination of the existing data therefore indicates that fracture mechanics techniques can be applicable, at least in the temperature and frequency ranges where crack growth is mostly cycle-dependent. It is possible, to some extent, to identify the cycle-dependent region since fracture in this region is mostly transgranular. For materials that are more ductile at high temperatures, J-integral may be better than ΔK or COD. Crack growth characteristics can be significantly altered when the time-dependent process is superimposed over the cycle-dependent process as discussed in the next section.

CREEP-FATIGUE CRACK GROWTH

With increasing temperature, decreasing frequency, or with hold-time at peak load, the contribution from time-dependent processes could become significant. In addition, increasing stress ratio, R, or increasing mean stress value could also increase the contributions from the time-dependent process. Prediction of crack growth in this range is important, since many service components are not subjected to pure cyclic nor pure static loads, but to a combination of them. Effects of hold-time on crack growth depends on the material and test conditions. These effects have been studied in many alloys [7,10,12,13,19,21,29,30,45-51], and they are summarized in Table 2 along with hold-time effects on uniformly stressed specimens. Note that hold-time effects observed in crack growth are also reflected in hold-time effects on fatigue life. This implies that the effects observed in fatigue life could be due to the effects in the crack propagation stage, although it does not rule out the possibility that the effects may be present in the nucleation stage as well.

Hold-time effects become significant when the microstructure becomes unstable at the test conditions. The instability in the microstructure could increase crack growth rates, as in Type 316 cold worked stainless steel, where dynamic recovery occurs, or decrease the rates, as in Alloy 718 at 760°C, where dynamic over-aging continuously alters the microstructure and the flow stress of the material. These factors have a significant impact in terms of the applicability of fracture mechanics to characterize crack growth. For such materials, crack growth surely depends on crack length, or more particularly on how long the material ahead of the crack tip is soaked at that temperature. Here the inapplicability of the fracture mechanics techniques is not due to the limitations in the parameters, but due to the

Table 2 Hold Time Effects on Fatigue Crack Growth and Life of Structural Alloys

Alloy	Condition	Temp (^oC)	Effects on Crack Growth	Ref	Temp (^oC)	Effects on Life	Ref
1 Cr-Mo-V	Bainitic	565	Significant increase Addition Rule does not work.	10	538	N_f signficantly decreases	45 46
2 1/2 Cr-Mo	Ferritic	510-593	Increased growth rates with decreased frequency from .66 to 0.066 Hz	7	593	N_f decreases, t_f increases	47
304 Stainless	Annealed	593	da/dN increases da/dt decreases	13	593-649	Decreased N_f, Increased t_f. Effects even in vacuum. Irradiation enhances the effect	48 49
	Annealed and aged	427-593	Negligible effect on da/dN	12 13 50	593	N_f nearly the same Irradiation causes hold time effects	48
	Cold worked	593	Large effect on da/dN Nearly the same on da/dt. - Mostly time dependent process	13 50	593	In Irradiated N_f decreases	48
	Cold worked and aged	593	Aging decreases the observed effect	50		-----	
316 Stainless	Annealed Annealed and Aged	593	Very small or no effect	50		Small effect	48
	Cold worked	593	Large increase in da/dN Increase even in da/dt	13	593	Small effect - more at high amplitudes	48
Alloy 718	Standard	538-760	Increase in da/dN - Largest effect at 650^oC Time dependent processes. Effect decreases at 760^oC.	19 21		-----	
Udimet 700	Standard	850	Depends on ΔK. Low ΔK - da/dN decreases High ΔK - da/dN increases Intermediate - Initial decrease and then increase	29	760	N_f decreases Compression more damaging	51
IN 100	Gatorized	650-734	da/dN increases for hold hold-time greater than 2 min.	30		------	

inability to account for the change in material properties ahead of the crack tip.

Hold-time effects observed in Udimet 700 provide further insight into crack growth characteristics under creep-fatigue conditions and provides an additional example where fracture mechanics techniques are inapplicable. Figure 6 shows the effects of 1-min hold-time on crack growth rates in Udimet 700 at 850°C [29]. For an initial ΔK of 40 MPa$\sqrt{m}$, the crack growth rates in a compact specimen decrease first with increase in ΔK and then increase as ΔK approaches the threshold ΔK for creep crack growth [52]. For higher initial ΔK values, crack growth rates increase continuously with ΔK and with hold-time on a da/dN basis. On the other hand, for still lower ΔK values close to 20 MPa$\sqrt{m}$, crack growth rates for 1-min hold decrease until the growth completely stops. This type of creep-fatigue interaction during crack growth occurs whenever there is a large disparity between the threshold values for cycle-dependent and time-dependent processes and when the applied stress intensities are below that of the threshold for the time-dependent process.

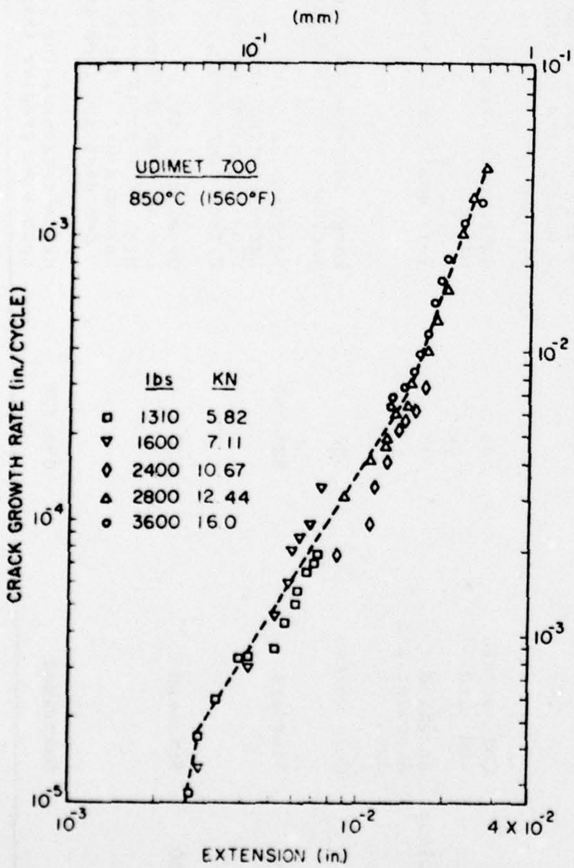


Fig. 5 Crack growth rates in Udimet 700 under load-controlled fatigue as a function of crack opening displacement

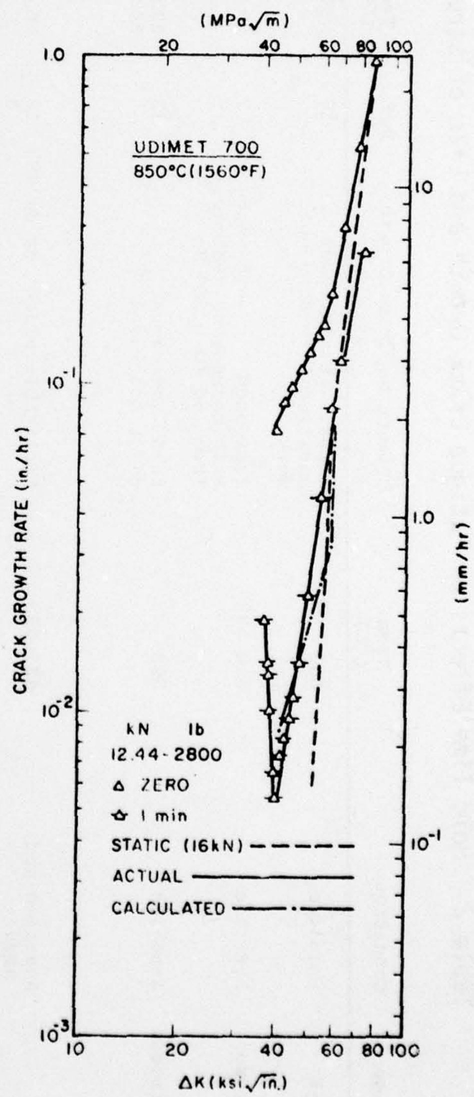


Fig. 6 Crack growth rates in Udimet 700 under creep-fatigue conditions

This can be seen clearly since in the limit of infinite hold-time (static load test) crack growth does not occur for ΔK less than the threshold ΔK for the time-dependent process. Such beneficial hold-time effects at low ΔK values were also observed in a low alloy steel [9], Alloy 718 at 760°C [21], and in directionally solidified carbide eutectics [35]. These effects become pronounced with increasing hold-time, with increase in temperature, and with decrease in flow stress of the material. This occurs generally in the transient regime, in terms of frequencies, hold-times, or temperatures, where crack growth changes from one process to the other, such as from cycle-dependent to time-dependent, or from one mechanism of crack growth to another. The inverted frequency dependence of crack growth rates that was predicted [21] and observed in a low alloy steel [9] is related to the above effects. Physically, the effects are related to the relaxation of the crack tip stress fields during hold times as discussed with reference to Fig. 2. Needless to say, crack growth rates in this regime cannot be characterized by any fracture mechanics parameter, including J-integral [28].

If crack growth rates increase continuously during a hold-time test, the fracture mechanics parameters are likely to be valid except when large scale deformation occurs which could reduce the stress concentration effect of the crack. This depends on the creep ductility of the material and test temperature. For superalloys which have relatively low ductility in the temperature range of application, fracture mechanics techniques are expected to be valid. If there are no interaction effects between the cycle- and time-dependent processes, the applicability of fracture mechanics techniques depends on the relative contribution from each process. Application of fracture mechanics concepts to characterize time-dependent crack growth will be addressed next.

TIME-DEPENDENT CRACK GROWTH

Creep Crack Growth

Creep crack growth is somewhat similar to sustained load cracking, as, for example, in Ti-alloys, but occurs at relatively high temperatures where time-dependent creep effects and enhanced environmental effects are superimposed. The relative contribution from each, however, depends on the material and its processing condition, flow stress, and temperature. Crack growth under static load at high temperatures has been determined in many alloys [8,10,13, 15,19,22,30,35,38,44,52-69] and is summarized in Table 3.

Fracture Mechanics Parameters

Similar to fatigue at high temperatures, various parameters have been tried to correlate crack growth data under static load. These include elastic stress intensity factor (K), J-integral, C*-integral, crack opening displacement rate ($\partial\delta/\partial t$), and nominal stress ($\sigma_n$). Some investigators showed that K is applicable, while others showed that other parameters are better than K in predicting creep crack growth. There seems to be some consensus, however, that for materials that have low creep ductility, K appears to be valid for a wide range of specimen geometries.

Stress intensity factor appears to be valid, at least for Type 316 cold worked stainless steel at 538°C, Alloy 718 in the range 538-650°C [19, 67], IN 100 in the range 650-704°C, and aluminum alloy at low K values at 150°C [55]. Stress intensity factor also appears to be better than nominal stress for Udimet 700 at 850°C [52], low alloy steels at 565°C [58], annealed 304 stainless steel at 538°C [63], and Alloy 718 at 650°C [67].

When creep rates are high, then the material ahead of the crack tip could deform and reduce the crack tip stress field. This could change the stress field to the extent that LEFM may not be applicable. In many cases

Table 3 Creep Crack Growth in Structural Alloys

Alloy	Temp (°C)	Geometry	Parameter	K Range (MPa$\sqrt{m}$)	Ref.
Al-Alloys					
RR 58	150	DCB, Con.K	K, C*	10-30	53,54
2219-T851	150	CT	K	15-45	55
Low Alloy Steels					
Cr-Mo-V	450-565	CC, SENT SENB, WOL CT	K, σ_N, $\frac{\partial\delta}{\partial t}$	10-90	8,10,44 56-60
Mild Steel	450	CT, SEN	K	30-50	61
Austenitic Stainless					
304	650	bar,plate	K, σ_N	8-15	62
308 weld	593	CT	K	40-60	15
316 sol., c.w.	538-740	CT, CC WOL, DEN	K, σ_N, $\frac{\partial\delta}{\partial t}$	8-40	13,63, 64
316 weld	538	WOL	$\partial\delta/\partial t$	38-50	65
Superalloys					
Alloy 718	425-760	CT, SEN, CC, Surface	K,J, C*	15-90	19,22, 66,67
Rene 95	535-760	CT	K	25-60	66
Astroloy	535-760	CT	K	25-60	65
Waspaloy	535-760	CT	K	25-50	38,66
Nimonic 115	704	bar	K	25-50	68
Discaloy	650	CT, CC	K, C*	30-80	69
IN 100 Gatorized	650-734	CT, CC	K	20-80	30
Udimet 700	850	CT	K	45-100	52
C 73 Carbide eutectic	750-950	SEN	K	30-80	35

correlations with K may still be obtained for one specimen geometry, but such correlation may not be valid for other geometries. In analogy with creep-rupture tests, nominal stress has been used to correlate crack growth data and is claimed to be better than K for low alloy steel at 550°C [8] and annealed 316 stainless steel at 740°C [64].

Haigh [65] argued that use of K is not justified since elastic displacements are small in comparison to the total. In such cases crack opening displacement rate, $\partial\delta/\partial t$ [44] was shown to correlate the data better than K.

In analogy with the fatigue results (see Fig. 4) J may be a better parameter to characterize creep crack growth, particularly when LEFM fails. Crack growth data of Udimet 700 [52] and Alloy 718 [67] were correlated with J-integral, but the results were not good. The large spread in the data obtained, however, may not be due to limitation of the parameter but due to experimental difficulties. It was shown with reference to fatigue [28] that

spurious values of J-integral can result if the load-displacement curve characterisitc of a given crack length is not properly defined. To determine the correct J-integral for creep crack growth, the specimen has to be unloaded and reloaded at each interval of crack increment to obtain the characteristic load-displacement curves for the given crack length. This, however, introduces a fatigue component into the crack growth. If the loading and unloading are done at slow rates, fatigue effects can be minimized and proper J-integral values could be determined.

Another energy-related parameter suggested for correlation of creep crack growth data is C*-parameter which is related to the crack tip stress-strain rate fields [69]. Following the procedures in Ref. [67, 69], C* can be determined for constant load or constant displacement rate tests. Landes and Begley [69] claim that crack growth rates correlate better with C* than with K in Discaloy. Figure 7 shows the data for Udimet 700 [52] in terms of K and C*. C* does not seem to be any better than K, at least for one geometry, although further evaluation is required using different geometries.

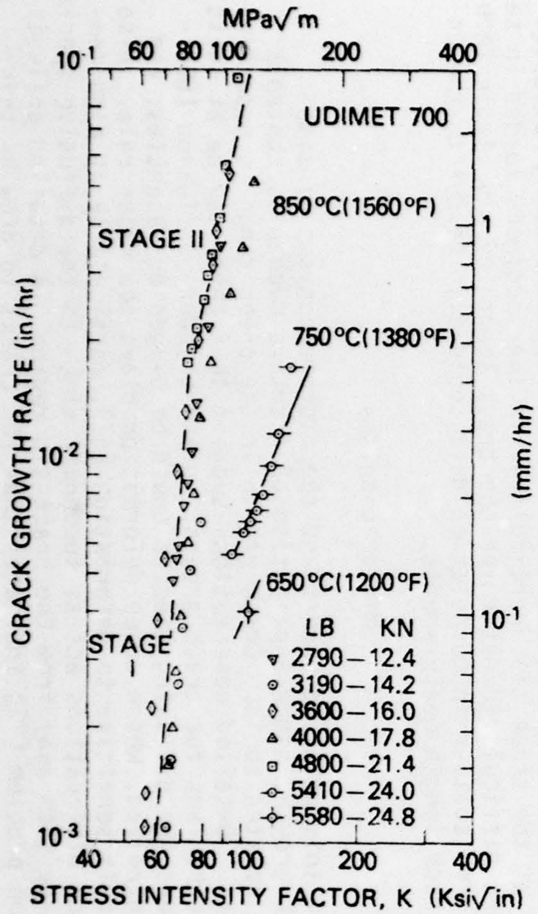
Turner and his coworkers [53,54,70] estimated C* parameter using non-linear bending theory. They show C* to be related to the strain rate sensitivity of the material, n, as well as to the product of load and $\partial\dot{\Delta}/\partial a$, where $\dot{\Delta}$ is the displacement rate, and a, as defined earlier, is the crack length. They show that C* correlates crack growth rate data reasonably well for aluminum alloy and low alloy steel. No attempt has been made to compare the estimated C* values with those determined experimentally [67,69]. Also, their analysis applies only when bending components are overwhelming as in DCB or CT specimens. C* parameter seems to be promising for creep crack growth, but additional work is required using different geometries and loading modes in different materials.

In extending the fracture mechanics concepts to creep crack growth, two problems arise which require careful consideration. This crack growth is highly sensitive to changes in microstructure [60,66,67]. Large differences in growth rates were observed in the same alloy which underwent the same heat treatment, but in two different batches [29]. This means that observed scatter, when comparing data from different studies, may not be due to the limitation in the fracture mechanics parameters. This also implies that to use fracture mechanics for design, close control of chemistry and processing treatments is required.

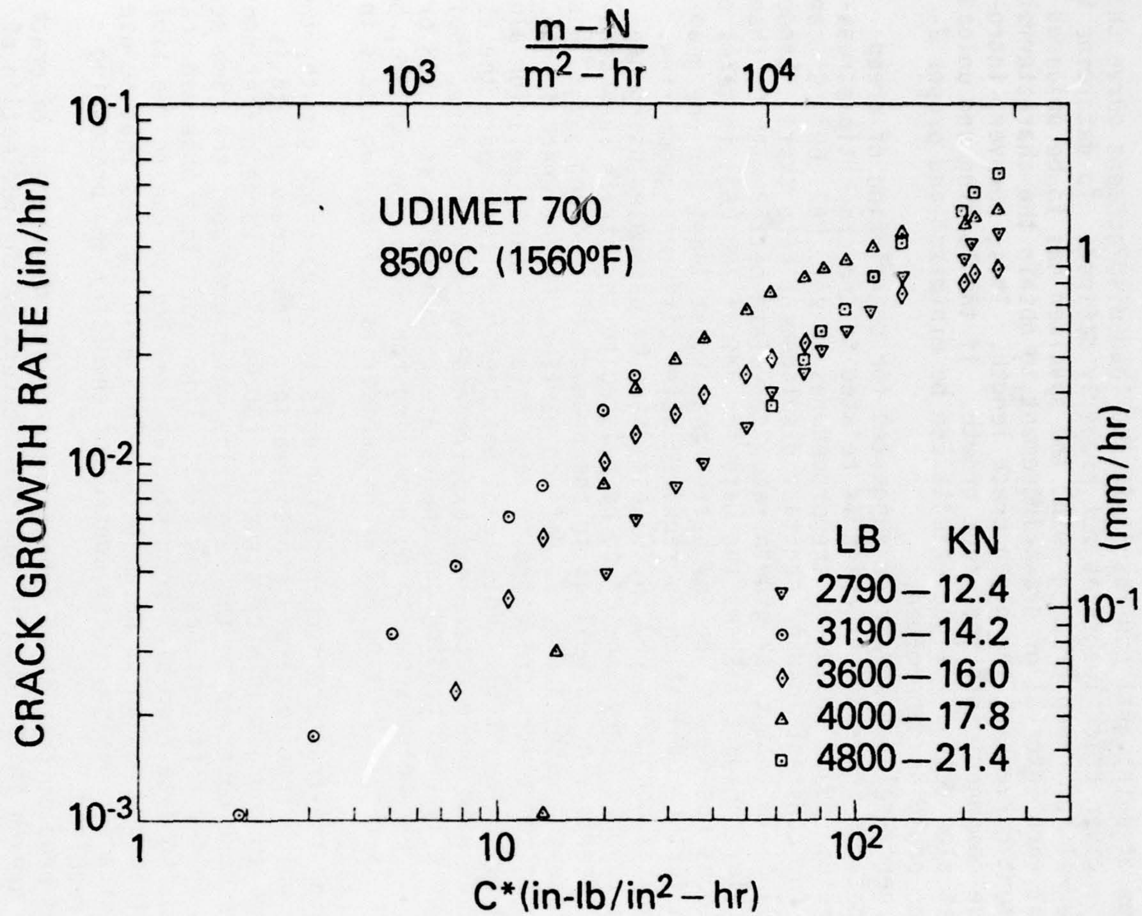
The second problem that is of concern here is the measurement of crack length. Crack growth in the ductile materials occurs by the nucleation of small cracks ahead of the main crack and joining of the most suitably oriented cracks with the main crack. When several cracks are being nucleated simultaneously at the crack tip, particularly in the specimens loaded in tension, it becomes difficult to determine the crack length by any viable technique. Therefore, scatter in correlation data could also be due to the uncertainty in crack length measurements.

Micromechanisms

It is useful at this point to understand the micromechanisms of time-dependent crack growth, since application of fracture mechanics concepts are intimately related to it. Crack growth in the creep range is essentially intergranular. Detailed observations suggest that there may be at least two distinct mechanisms for crack growth: grain boundary diffusion (GBD)-controlled, where the diffusing species could be oxygen or vacancies; and deformation-controlled, where creep deformation plays the major role. Also, the GBD process is sensitive to deformation [67], particularly at high temperatures, since dislocations act as secondary sinks to the diffusive species and thus disperse them away from the crack tip region. A detailed analysis of the diffusion problem [71] indicates that for cracks to grow by this



(a)



(b)

Fig. 7 Creep crack growth rates in Udimet 700 as a function of (a) Stress intensity factor and (b) C* - parameter.

process, $D_g\delta > 2D_b r_p$, where D_g and D_b are grain boundary and bulk diffusion coefficients, δ is the grain boundary thickness, and r_p is the plastic zone size. Since creep deformation is not a prerequisite for GBD controlled crack growth, cracks can grow rapidly, especially in high strength materials, before creep relaxations can set in. Fracture mechanics concepts should be readily applicable for this crack growth. With increase in temperature, deformation rates on one side and bulk diffusion rates on the other increase rapidly to induce deformation-controlled crack growth processes. Crack growth in this regime occurs by the nucleation of voids or cracks ahead of the main crack and their subsequent coalescence. Since extensive deformation could reduce the stress concentration effect at the crack tip, fracture mechanics may have limited application in this range. Fracture mechanism maps were developed based on the above concepts [71] which show the temperature and stress intensity ranges where GBD and deformation-controlled mechanisms occur.

For time-dependent crack growth, applicability of fracture mechanics depends on the micromechanism of crack growth. For GBD controlled process, which is likely to occur in the temperature range 0.4 to 0.7 T_M, fracture mechanics techniques are applicable. For deformation-controlled crack growth, which is likely to occur at higher temperature or in relatively ductile materials, the application of the fracture mechanics techniques may be limited. The techniques can be applicable until the deformation becomes significant which reduces the stress concentration effect of the cracks.

SUMMARY AND CONCLUSIONS

Application of fracture mechanics concepts to characterize crack growth at high temperature under cyclic, static, and combined loads has been reviewed. When crack growth is cycle-dependent, linear elastic fracture mechanics is applicable to a ΔK range where plane strain conditions exist; but this range is not necessarily related to that given by the ASTM criterion for plane strain for fracture toughness. It may be necessary to develop alternate plane strain criteria for fatigue and creep which depend on the micromechanisms of crack growth. Fracture mechanics concepts could be utilized to a limited extent for time-dependent crack growth. Non-linear parameters may be more appropriate for this. Extensive work using different specimen geometries and loading for different materials is required before the limitations of the fracture mechanics concepts to time-dependent crack growth is fully ascertained. Application of the concepts to crack growth under combined creep-fatigue conditions depends on several factors, particularly those which influence creep-fatigue interactions. Microstructural instabilities during crack growth would also limit their application. In addition, the applicability of fracture mechanics concepts are shown to be intimately related to the micromechanics of crack growth, since the time-dependent processes are very sensitive to the variations in microstructure.

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